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THE NATURE OF RESIDUAL STRESS AND ITS MEASUREMENT.(U)
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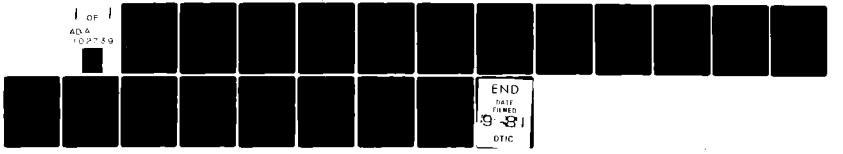

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THE NATURE OF RESIDUAL STRESS AND ITS MEASUREMENT

BY

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THE NATURE OF RESIDUAL STRESS AND ITS MEASUREMENT

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ABSTRACT

The origins of residual stress and changes during fatigue are discussed. A new mechanism for fading is proposed. Practical (destructive and non-destructive) methods for measuring this stress are critically reviewed. Each technique has major problems requiring further study, but acoustic, magnetic and x-ray methods are all poised for more widespread use.

INTRODUCTION

This is an important and timely conference because we are on the verge of making important steps in measuring residual stress, in standardizing the procedure and in understanding the role this stress plays in many processes, especially how it changes during use. This week, in these pleasant surroundings, we have the unusual opportunity to hear from people around the world about this subject and the responsibility to discuss key issues openly and vigorously; at such a meeting our participation is more vital than the presentations. In this introduction to the conference we hope to emphasize some of the issues. Fortunately, we do not also have to provide the answers.

THE ORIGINS OF RESIDUAL STRESS

We can define residual stress as the self-equilibrating internal stress existing in a free body when no external tractions are applied. At equilibrium the integral of this stress in the volume of the body

must be null, and, as well, the integral over any plane through the specimen. In Fig. 1, two parts of such a body that we will refer to as "bulk" and "near surface" regions are shown separated. Suppose that for some reason the near-surface is elongated plastically. Then it is compressed elastically to join it to the bulk and released. The bulk puts this region under compression, while the near-surface exerts tension on the bulk. Bending can result, depending on the magnitude of the stress and relative thicknesses of the layers. In almost every real situation we can think of residual stress arises in this manner.

Residual stress can arise, for example, when a material is subjected to heat treatment or machining. Consider first a material that undergoes no change in crystal structure during heat treatment. If aluminum is cooled quickly from a heat treatment, the surface and the interior contract at different rates, as illustrated in Figure 2a. At some time, A, this difference, coupled with low material yield strength associated with the high temperature, induces plastic flow or permanent yielding. The surface region, which, because of the temperature gradient, contracts on cooling more than the interior, is extended by the interior and vice versa. (Note the increase in length in the surface, A in Figure 2a.) This is a real effect; for iron-base materials the product of Young's modulus and coefficient of expansion yields a stress of 3.5 MPa (.5 ksi) per°C difference in temperature between two such regions.

On continued cooling to room temperature, point B, the surface regions have been extended relative to the interior and consequently end up in compression. Residual compressive stress in the surface must be overcome by the applied load to initiate cracks, and thus the presence of surface compressive stress is a highly favorable condition.

Heat treatment does not always produce surface compressive stress. If a material undergoes a phase transformation, as in the hardening of steel associated with martensite, the local yielding is essentially masked by the volume expansion associated with the

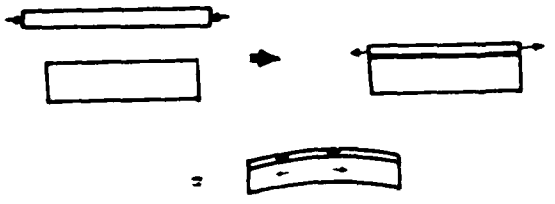
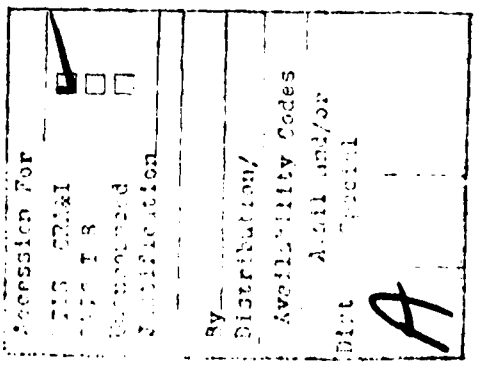
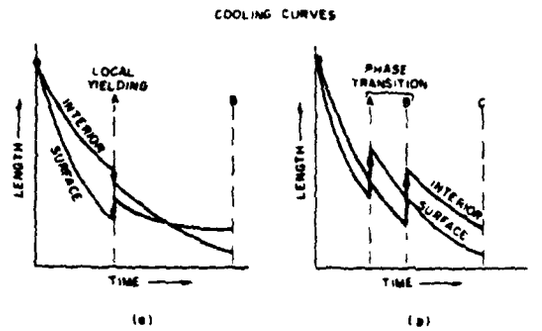
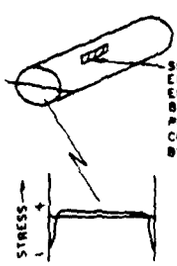


Fig. 1. If the near surface region is longer than the bulk compressive residual stress occurs at the surface, tensile residual stress in the bulk.



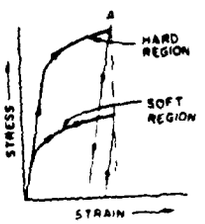


Schematic cooling curves during a heat treatment showing the difference in contraction of the surface and interior. In (a) there is no phase transition, whereas one occurs in (b).

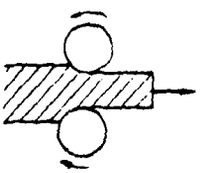


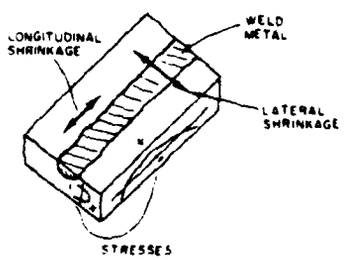
(c) — Surface deformation by peening elongates surface regions

(d) — The harder surface regions contract more than release on the load at A1 than the softer interior



(e) — Friction at the rolls in a rolling mill tends to extend surface regions more than the interior





(f) — Stresses due to welding

Figure 2

austenite to-martensite phase change. The result is illustrated schematically by the cooling curve in Fig. 2b. At temperature A the surface region transform to martensite and expands, since it reaches the transformation temperature first. The interior composed of low strength austenite deforms plastically to accomodate this change. At B the interior transforms to martensite producing an expansion which is resisted by the high strength martensite surface. At C, near room temperature, the surface is thrown into tension by the interior, producing surface residual tensile stress, which can contribute to crack initiation and propagation. The origin of this stress is much more complex in a steel that is case hardened and undergoes different transformations at different temperatures during cooling. In Prof. Ericsson's chapter this is discussed in detail.

Stress relief annealing at moderate temperatures is often employed to allow local yielding to occur, thereby minimizing or eliminating residual stress. However, care is needed because even differences in the coefficients of expansion between the carbide and ferrite phases in steel can lead to significant stress if the parts are not cooled slowly after this treatment.

Another way to produce compressive stress in the near-surface region is to shot peen the surface. In this process, high velocity shot causes local plastic yielding in the surface, which is extended relative to the interior. The interior acts to constrict the surface, resulting in high, local compressive residual stress in the surface, balanced by tensile stress within the interior, Figure 2c. Prof. Wohlfart covers this topic, as well as Mr. Canmett.

In fact even a tensile extension of a specimen into the plastic region can produce stress. If the surface is harder than the interior because of defect pileup occurring during plastic extension, then on release of the load (at A in Figure 2d), elastic recovery leaves the surface shorter than the interior, resulting in surface tensile stress. (The reverse occurs when the surface is softer than the interior.) In a forming operation, such as rolling, the surface can be extended more than the interior due to friction at the rolls, as illustrated in Figure 2e, resulting in compression in the surface. The magnitude of the stress is a function of the thickness of the piece, the roll size and the degree of reduction.

Another important cause of residual stress is welding, as illustrated in Figure 2f. Contraction of molten weld metal during solidification is resisted by colder surrounding metal resulting in the stresses illustrated in the figure. (For further information on stresses in this process see ref. 2.) Prof. Masubuchi reviews this matter in this volume. From these examples (see also refs. 3-5) it is clear that residual stress in materials arises not only in processing, but also in use. In Fig. 3, we show some examples of how even ordinary surface preparation can produce a large stress.

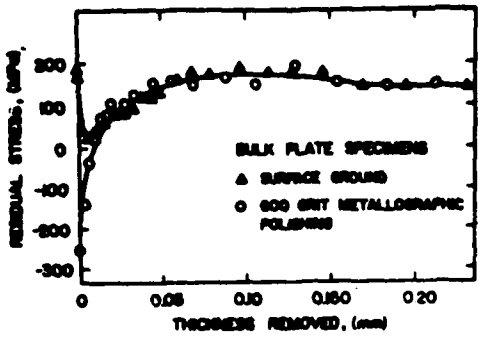


Fig. 3. Residual stress in an HSLA steel due to surface preparation. From ref. 6.

Unfortunately, in too many cases residual stress is ignored, or it is assumed that additional treatment has either eliminated it or introduced compressive stress. To further complicate the issue, not only will "macrostress" develop in different regions of the same piece, but "microstress" can arise in microscopic regions, such as between the phases of a multiphase material. The magnitude of this stress can be a significant portion (half or more) of the ultimate tensile stress of the annealed material. Prof. Hauk presents information on this matter.

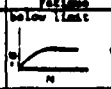
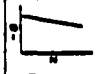
CHANGES IN RESIDUAL STRESSES DURING USE

Many of our speakers are concerned with this topic. As an example we will consider the fatigue process, first reviewing what is known, and then proposing a qualitative rationale for the observed behavior.

In low cycle fatigue of a rolled HSLA steel, Quesnel et al.[6] found that residual stress changed sign each half cycle, being opposite in sign to the sign of the applied load before load release. In Table I we summarize investigations on plain carbon steels in high-cycle fatigue. A clear pattern emerges. Below the fatigue limit with annealed specimens, compressive stress forms and saturates. In one study[11] this stress was found to occur only in deformation markings. Above the fatigue limit, this residual stress develops and then decreases until fracture. For specimens that are shot peened, relaxation occurs at all stress levels and is most rapid in the early stages of fatigue. In tension-tension fatigue of a shot peened piece the stress can even reverse sign and become tensile, a clearly detecterous effect. Much more work is needed with various load histories as most previous work is in bending or torsion. However we consider here the data currently available.

Fading is often attributed to microcrack formation, although there is as yet little evidence for this. Another interpretation

Table 1. Summary of Some Studies of Changes in Residual Stresses: Plain Carbon Steels During High Cycle Fatigue

Author	Material & Composition.	Heat Treatment & Mechanical work.	R ratio & Type of Fatigue Test.	Applied Stress (± Fat. limit)	Direction of residual stress	Results Fatigue below limit	Results Fatigue above
Taira & coworkers(7-10)	.07% C,.16%C, .28% C	Annealed	R = -1 Bending	.83-1.19	Axial and transverse in some cases		
Taira & coworkers(7,8)	"	Cold-Worked	"	"	"		
McClinton & Cohen (11)	.45% C	Annealed	R = 1 Axial pull-pull	Uncertain	Axial		
Torassem et al. (12)	.16% C	Tufftrided	R = -1 Bending	1.10	Axial		
Ericsson (13)	.45% C	↑	↑	↑	Axial	Compressive stresses reversed sign and became tensile.	
McClinton & Cohen (11)	0.40% C	Normalized & peened	R = 1 axial pull-pull	at or above endurance limit	Axial	Compressive stresses reversed sign and became tensile.	
Pattinson & Dugdale (14)	0.17% C	Normalized and straightened	Reverse bending, strain control	0.3	Axial	Fading	
Rodenn(15,16)	0.17%C	Annealed & shot peened	R = 1 bending	above endurance limit	Axial	Fading in two stages, rapid at first.	
Syren, Wahlfart & Mecherauch(17)	0.45% C	Quenched in oil, shot peened	bending	above endurance limit	Axial	Subsurface cracking in peened specimens where stress profile changed sign.	

by Taira's group[18,19] is that the residual stress forms in annealed specimens due to elongation of the near-surface region from excess vacancies formed during cycling. This continues until work hardening saturates, at which time the maximum compressive stress occurs. Additional cycles produce deformation of deeper layers, resulting in relaxation. Thus, relaxation should occur only in tensile cycles. However no relaxation is observed in this case; see Table I. James[20] suggest that stress can relax due to microplasticity in the near-surface region (see the chapter by James). As the surface is initially in compression, the relief would occur only in a compressive half cycle. However peened specimens do show stress relief in tensile cycling, Table I.

As an alternative to these theories, we propose the following qualitative rationale. Formation of compressive stress at the surface can occur only by:

1. Micro-plastic elongation in the near-surface region, with respect to the bulk; the bulk is stretched elastically, and places the near-surface in compression.

2. Alternatively, micro-plastic contraction of the bulk will cause the near-surface region to be placed in compression.

In *tension-tension fatigue*, compressive residual stress develops. As no contraction of the bulk is possible in this case, we can conclude that the dominant process causing the compressive stress is(1). Stress develops in the tensile portion of a cycle.

A maximum in the residual stress occurs during bending fatigue above the endurance limit, followed by fading. Fading can occur if:

1) the bulk elongates plastically or,

2) the surface contracts plastically.

As no relaxation appears to occur in tensile cycling of annealed specimens, we can conclude that relaxation occurs in a compressive half cycle and follows (2); surface elements are shortened.

The maximum in residual stress may occur because the surface can sustain only a certain plasticity without void formation, or because of local work hardening.

After shot peening, the observation that relaxation occurs in tensile cycling leads us to believe that the bulk is elongating plastically more than the (hardened) surface.

These simple qualitative ideas seem adequate to explain the

known results and suggest a number of interesting experiments. For example, residual stress should be examined after each half cycle of fatigue, to see when formation and relaxation actually occurs.

Residual stress should be measured in the axial and transverse directions, and in a direction normal to the surface. Such results might lead to an understanding of why in some cases the sign of the residual stress is not important in fading[21], whereas in other cases it is clear that the algebraic sum of applied and residual stress is controlling.[22]

THE MEASUREMENT OF RESIDUAL STRESS

An entire session is devoted to this subject, so here we only briefly review procedures, emphasizing especially their limitations, as a basis for discussion in subsequent sessions.

1) Destructive Methods

One popular procedure is hole drilling or disection. The relief of stress distorts the region around the hole and the stress is obtained from measurements of this distortion with strain gauges. Care is needed to avoid producing large stress in the drilling operation. Also, the hole itself is a stress concentrator, and this can lead to unwanted local plastic deformation, contributing to the distortion.

Another common method is to remove material by boring or electropolishing and to determine the stress from measurements of strain on the surface opposite to the one where material is removed.[23] In applying such methods, problems can also arise. For example, if a soft material has a shallow heavily deformed layer, removing material may lead to plastic deformation of the interior. Furthermore, the principal stress axes are not necessarily in the surface, although this is usually assumed.

2) Non-destructive Methods

a. Acoustics[24]

The basic idea behind this method is that most solids are anharmonic; when a stress is applied there is a change in the elastic constants. Therefore the velocity of wave propagation is altered by stress. Popular methods of analyzing stress include acoustic birefringence involving the measurement of velocity or more precisely the transit time or phase difference in two directions with respect to the stress system. By varying the frequency, the depth of sampling can be varied. Surface elastic waves are also utilized. All such techniques suffer

from the fact that the effective higher-order elastic constants are sensitive to microstructure, texture, and in some cases to plastic deformation. Nonetheless with careful calibration this method should be receiving more attention than has been the case to date. Why is this so?

3) Magnetic Methods[24]

The magnetization curve of a ferromagnetic material is not as smooth as it is commonly drawn. The discrete motions of magnetic domains cause small sharp changes in magnetization as the field is applied, as shown in Fig. 4, and these can be employed to induce current in a nearby pick-up coil. This "Barkhausen noise" is proportional to stress because local strain centers impede domain motion, altering the number and amplitude of the oscillations in magnetization. As this explanation implies, local variations in dislocation density and microstructure can alter the signal as well as stress. Also, the signal saturates at ~100 ksi (tension or compression). Despite these problems, with calibration, this technique is promising for field measurements on steels.

4) Diffraction[25,26]

In Fig. 5a the incident beam diffracts radiation from grains with planes parallel to the surface to form a diffraction peak, recorded by a detector moving around the specimen. If there is, say, a compressive residual stress parallel to the surface, due to Poisson's effect, these planes have a larger spacing ("d") than an unstressed specimen, and the peak occurs at a lower scattering angle, according to Bragg's law. If the specimen is tilted with respect to the incident beam, Fig. 5b, other grains diffract. These have planes more nearly perpendicular to the stress, and have their spacing reduced. The peak is shifted to higher angle. These changes in spacing are a measure of the residual strain. By combining measurements at two or more tilts, the strain can be obtained without knowledge of the unstressed interplaner spacing. In the last decade there have been numerous developments in this technique which we list here briefly (see ref. 25,26 for further

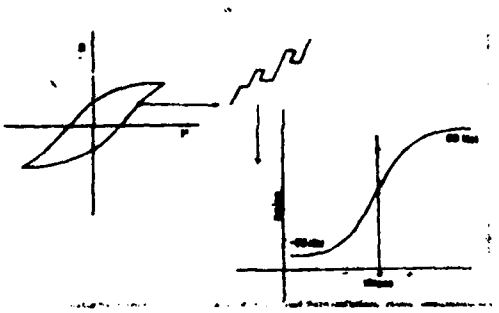
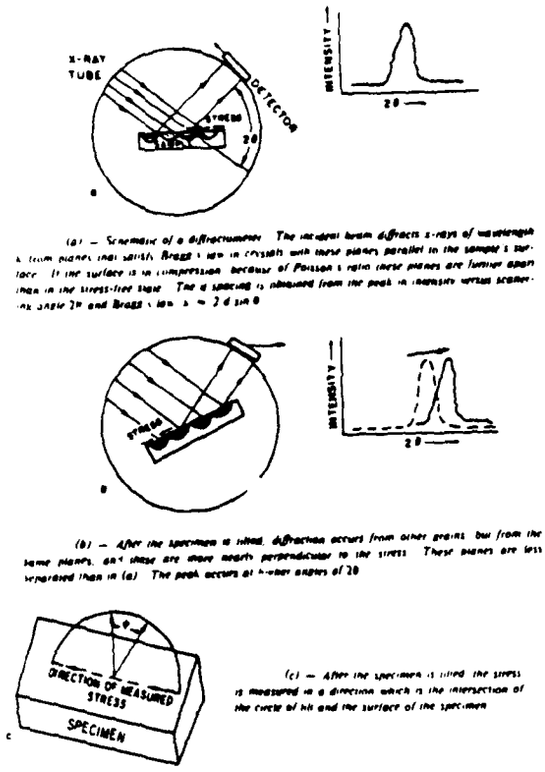


Fig. 4. Left, any segment of a B-H curve actually consists of discrete changes in B (magnetization) which can be sensed with a pick-up coil. Right: This "noise" is related to stress.

Figure 5



(a) — Schematic of a diffractometer. The incident beam diffracts x-rays of wavelength λ from planes that satisfy Bragg's law in crystals with these planes parallel to the sample's surface. If the surface is in compression because of Poisson's ratio these planes are further apart than in the stress-free state. The d spacing is measured from the peak in intensity versus scattering angle 2θ and Bragg's law λ = 2d sin θ

(b) — After the specimen is tilted, diffraction occurs from other grains but from the same planes, and these are more nearly perpendicular to the stress. These planes are less separated than in (a). The peak occurs at higher angles of 2θ

(c) — After the specimen is tilted the stress is measured in a direction which is the intersection of the circle of tilt and the surface of the specimen

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details, or the chapters by Profs. Hauk and Krawitz).

Developments in Equipment

1. Use of a parallel rather than a diverging beam, which minimizes angular shifts due to incorrect positioning of a sample on a diffractometer.

2. Tilting about a horizontal axis instead of the axis vertical in Fig. 5. (The tilt shown in Fig. 5 can cause the beam to be blocked with complex parts.)

3. Automation, to produce results to an operator-specified precision.

4. The use of position sensitive detectors which see the entire diffraction peak at once without motion of the detector. The time of measurement is reduced by factors of 3-100 depending on the desired precision and

the width of the peak.

5. Portable units capable of measurements in the field or in a factory environment, and in times of 20-30 seconds or less. This equipment is usually employed to examine residual stress. However, there are numerous practical situations where a known load or torque is applied in a device. With the availability of field instruments, the same techniques can be employed to monitor changes in these applied loads.

Developments in Technique

1. Formulae available for calculating geometric and statistical errors.

2. Recognition of the importance of often making measurements of "d" at several ψ values (see items below for reasons) and that such measurements can be made in the same total time as the two-tilt procedure.

3. The "d" spacing vs $\sin^2\psi$ (where ψ is the tilt of the specimen) should be linear, according to the theory of this measurement and the stress is obtained from the slope. However, oscillations sometimes occur in "d" vs $\sin^2\psi$ for textured materials. These oscillations are due primarily to elastic anisotropy, and can be minimized by the proper choice of reflection.[27]

4. The usual x-ray wavelengths employed (Cr K_α, CoK_α) penetrate a few tens of microns. If there is a steep gradient, stresses normal to the surface are included. These lead to a gradual curvature in "d" vs $\sin^2\psi$, which can be employed to measure this normal stress, and to estimate the gradient without removing layers. Alternatively, with different wavelengths different depths can be sampled.

5. The presence of shear stresses normal to the surface in such gradients can be detected because "d" vs $\sin^2\psi$ has different curvature for $\pm\psi$. The entire three dimensional stress gradient can be obtained.[28] This can be especially important in studies of wear, or of surface finishing (grinding, etc.).

6. The measurement of stress in the bulk is possible using high energy x-rays or neutrons. The procedure is sketched in Fig. 6. By moving the receiving slit different volumes can be sampled. This method is

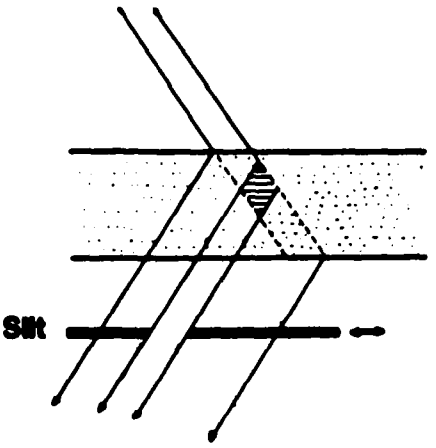

Fig. 6. Neutron scattering - by moving the slit as shown the stress in different volumes may be measured.

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covered by Prof. Krawitz. It is not as impractical as it may seem. Dr. W. Yelon of University of Missouri-Columbia is developing packaging for high intensity x-ray sources. Such sources would require using a high-order diffraction peak to keep the measurement at high scattering angles which is required (for accuracy) and it should be kept in mind that the intensity of a diffraction peak from a polycrystalline specimen falls approximately proportional to the square of the energy due to the Debye-Waller factor; thus intensities will be low with x-rays, and the time for measurement long.

Despite this hoard of successes, three main problems remain. For accuracy (not precision) the actual x-ray elastic constants must be measured for the diffraction peak employed, by applying known stresses. This is, in effect, a calibration and (as with the other methods) is not a special problem. It is troublesome, however, that there are reports in the literature of changes of 20-40 pct in these constants with plastic deformation. In a practical sense this is still not a problem because the calibration should be made on a specimen as identical to the real one as possible. However, it would be best if the reason for these changes were understood. Secondly, even if measured elastic constants are employed significant differences in the stress are obtained with different diffraction peaks. Why?

A final problem is that if the material is quite coarse grained, as in a casting or weld, the diffraction peak can vary considerably over the area of interest, simply because only a small number of grains is diffracting. In the appendix we show one way

this problem can be cured, by simply measuring in one grain..

The diffraction method has one feature which is simultaneously an advantage and a disadvantage. The averaging distance is the size of a subgrain, whereas the other methods average over much larger distances. In a multi-phase material the stresses in one phase may not sum to zero over the sample volume, because part of the load is supported by the other phases. The stress in these other phases can also be measured, and more work of this kind is sorely needed.

Discussions are underway in the U.S. to develop standard procedures for all these techniques. Undoubtedly this will be helpful, because much can be learned about the remaining problems in developing these standards, and in using them.

ACKNOWLEDGEMENTS

The authors take special pleasure in acknowledging the long-standing support of our efforts in this area by the U. S. Office of Naval Research, particularly Dr. B. A. MacDonald.

APPENDIX

We consider the determination by diffraction of the three-dimensional stress tensor for a single crystal or grain in a coarse grained specimen. The orthonormal co-ordinate system is illustrated in Fig. A-1; the S_i define the sample, with S_1, S_2 in the surface and S_3 normal to it, whereas the measurement for an hkℓ plane is made along L_3. The axis L_2 is in the plane defined by S_1 and S_2 and the axis L_1 is the vector cross product L_2 and L_3. Strains in the L_i axial system will be primed, those in the P_i, unprimed. A normal to a plane is uniquely defined by the angles χ, Φ. The direction cosines that link the axes S_i to L_i, (that is the axes L_i in terms of S_i) form a matrix:

	S_1	S_2	S_3
$(L_1)_{\chi\psi}$	$\cos\psi\cos\chi$	$\sin\Phi\cos\chi$	$\cos(90^\circ+\chi)$
$(L_2)_{\chi\Phi}$	$\cos(90^\circ+\Phi)$	$\cos\Phi$	$\cos 90^\circ$
$(L_3)_{\chi\Phi}$	$\cos\Phi\cos(90^\circ-\chi)$	$\cos(90^\circ-\Phi)$	$\cos\chi$

(A-1)

As the strains (ϵ_{ij}) are a rank 2 tensor:

$$\epsilon_{ij} = a_{ik}\, a_{j\ell}\, \epsilon_{k\ell}. \qquad \text{(A-2)}$$

Therefore:

$$\frac{\Delta d}{d} = (\epsilon'_{33})_{\chi,\Phi} = \epsilon_{11}\cos^2\Phi\sin^2\chi + \epsilon_{12}\sin^2\Phi\sin^2\chi + \epsilon_{13}\cos\Phi\sin 2\chi$$
$$+ \epsilon_{22}\sin^2\Phi\sin^2\chi + \epsilon_{23}\sin\Phi\sin 2\chi + \epsilon_{33}\cos^2\chi. \qquad \text{(A-3)}$$

With six or more independent reflections for which χ, Φ can be calculated (or experimentally measured since stress may cause them to change), the six unknown ϵ_{ij} can be obtained. These can then be referred to the crystal axes using Eqn. A-2 and the direction cosines between the axes P_i and the crystal basis. The stresses in the crystal axial system can be obtained from:

$$\sigma_{ij} = C_{ijk\ell}\,\epsilon_{k\ell}. \qquad \text{(A-4)}$$

These may be referred back to the axes P_i by the inverse of Eqn. A-2 (with σ_{ij} replacing ϵ_{ij}).

In many cases $\sigma_{13} = \sigma_{23} = \sigma_{33} = 0$, that is, there are stresses only in the surface plane. We assume that the co-ordinates S_i are along the principal axes. Then:

$$\epsilon_{33}' = S'_{33k\ell}(\sigma'_{k\ell}), \qquad \text{(A-5)}$$

and to obtain the S' from S'' in the crystal axial system:

$$S'_{ijk\ell} = a_{im}a_{jn}a_{ko}a_{\ell p}S''_{mnop}. \qquad \text{(A-6)}$$

Also:
$$\sigma_{ij} = a_{ik}a_{j\ell}\sigma_{k\ell}. \qquad \text{(A-7)}$$

Employing Eqn's A-5 to A-7 for this special case:

$$\epsilon_{33}' = A\sigma_{11} + B\sigma_{22}, \qquad \text{(A-8)}$$

where: $A = S'_{31}\cos^2\Phi\cos^2\chi - (S'_{36}/2)\sin 2\Phi\cos\chi + (S'_{35}/2)\cos^2\Phi\sin 2\chi$
$$+ S'_{32}\sin^2\Phi - (S'_{34}/2)\sin 2\Phi\sin\chi + S'_{33}\cos^2\Phi\sin^2\chi, \qquad \text{(A-9)}$$

and: $B = S'_{31}\sin^2\Phi\cos^2\chi + (S'_{36}/2)\sin 2\Phi\cos\chi + (S'_{35}/2)\sin^2\Phi\sin 2\chi$
$$+ S'_{32}\cos^2\Phi + (S'_{34}/2)\sin 2\Phi\sin\chi + S'_{33}\sin^2\Phi\sin^2\chi.$$

As a practical example, consider a Si wafer with a (111) plane as the surface. The stress in such wafers can be important in processing electronic devices. A stereographic projection is shown in Fig. A-2. The "d" spacing of the 513 reflection is ~ 0.918 Å and, with CoK_α, $2\theta = 154^\circ$, so that the diffraction peak occurs at high angles where precision is good. In fact this angle is very similar to the value employed for steel. The 513 and 153 reflections are 57.1° apart and a simple tilt suffices to cause either reflection.

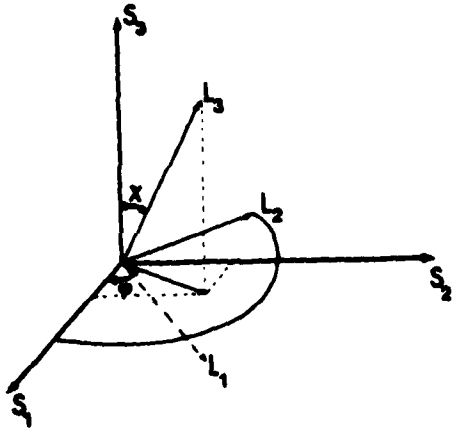

Figure A1. The axial system.

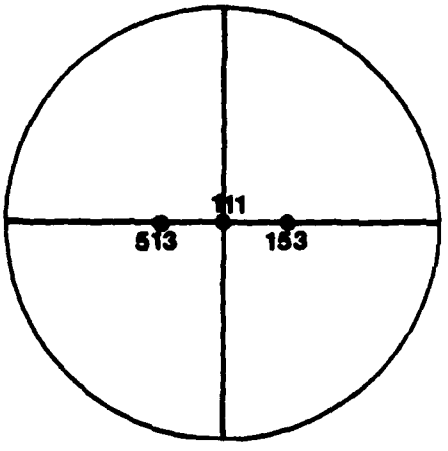

Figure A2. A cubic crystal with [111] normal to its face.

REFERENCES

1. W. M. Baldwin, Jr.,"Residual Stresses in Metals" Proc. ASTM. 4a: 539 (1949).
2. V. Papazoglou and K. Masubuchi, Final Technical Report, ONR. Contract No. N00014-75-C-0469.
3. G. M. Rassweiller and W. L. Grube (eds.), "Internal Stresses and Fracture in Metals", Van Nostrand Co., New York (1959).
4. O. J. Horger, "Residual Stresses", Metals Eng. Design, ASME Handbook, 2nd ed. McGraw Hill, New York (1965).
5. F. A. McClintock and A. S. Argon, "Mechanical Behavior of Materials", Addison - Wesley, Reading Mass. (1966) pp. 420-442.

6. D. Quesnel, M. Meshii and J. B. Cohen, "Residual Stresses in High Strength Low Alloy Steel During Low Cycle Fatigue", Mat. Sci. & Eng. 36(2): 207 (1978).
7. S. Taira and Y. Murakami, "On the Contribution of Residual Stress to Coating", Proc. 4th Japan Congress on Testing Materials, p. 5 (1961). Kyoto, Soc. Mat. Sci., Japan.
8. S. Taira and K. Honda, "X-ray Investigation of Fatigue Damage of Metals", Proc. 5th Japan Congress on Testing Materials, p. 8 (1962). Kyoto, Soc. Mat. Sci. Japan.
9. S. Taira and K. Hayashi, "X-ray Investigation of Fatigue Fracture of Notched Steel Specimens", Proc. 9th Japan Congress on Testing Materials, pp. 1-6 (1966). Kyoto, Soc. Mat. Sci., Japan.
10. S.Taira, T. Abe and T. Ehino, "X-ray Study of Surface Residual Stress Produced in Fatigue Process of Annealed Metals", Bulletin of JSME, pp. 947-957, (1969).
11. M. McClinton and J. B. Cohen, "Changes in Residual Stress During Fatigue of Normalized and Peened 1040 Steel", to be published.
12. M. Terasawa, Y. Yoshioka and K. Asami, "Studies on Fatigue Fracture of Tufftrided Steel", Proc. 11th Japan Congress on Materials, p. 39 (1968). Kyoto, Soc. Mat. Sci., Japan.
13. T. Ericsson, "Residual Stresses and Fatigue in Steel", Proceedings of Fatigue (Fundamental and Applied Aspects) Seminar. Rimyorsa, Sweden (1977).
14. E. J. Pattinson and D. S. Dugdale, "Fading of Residual Stresses Due to Repeated Loading", Metallurgica, 66: 228 (1962).
15. S. Kodama, "The Behavior of Residual Stress During Fatigue Stress Cycles", Proc. Int. Cong. Mech. Behavior of Materials, pp. 111-118 (1972). Kyoto, Soc. Mat. Sci., Japan.
16. S. Kodama, "On the Decrease of Residual Stress Due to Cyclic Stress", Paper V/4418, 3rd Int. Cong. on Fracture, Munchen (1973). Vereir Deutscher Eisenhutterleute.
17. B. Syren, H. Wohlfart and E. Macherauch, "The Influence of Residual Stresses and Surface Topography on Bending Fatigue Strength of Machined CK45 in Different Heat Treatment Conditions", Proc. 2nd Int'l. Congr. Mech. Behav. Materials, Boston, p. 212 (1976).
18. S. Taira, T. Abe, and T. Ehiro, "X-ray Investigation of Surface Residual Stress Produced in Fatigue Process", the 11th Japan Congress on Materials Research, pp. 32-38 (1969). Kyoto, Soc. Mat. Sci., Japan.
19. S. Taira, T. Abe and T. Ehiro, "X-ray Studies of Surface Residual Stress Produced in Fatigue Process of Annealed Metal" Bull. JSME, 53:947 (1969).
20. M. R. James and W. L. Morris, "The Relaxation of Machining Stresses in Aluminum Alloys During Fatigue", Residual Stress for Designers and Engineers, ASM, Metals Park, Ohio p. 189 (1981).

21. J. M. Potter and R. A. Millard, "The Effect of Temperature and Load Cycling on the Relaxation of Residual Stresses", Adv. in X-ray Analysis, 20:309 (1977).
22. D. Rosenthal: "Influence of Residual Stress on Fatigue","Metal Fatigue" (eds: G. Sines & J. L. Waisman) McGraw-Hill, New York, p. 170 (1959).
23. K. Heindlhofer,"Evaluation of Residual Stresses",McGraw-Hill, New York (1948).
24. M. R. James and O. Buck, "Quantitative Nondestructive Measurements of Residual Stresses",CRC Critical Reviews in Solid State and Materials Sciences, 9: 61 (1981).
25. M. R. James and J. B. Cohen, "The Measurement of Residual Stresses by X-ray Diffraction Techniques", Exp. Methods of Mat'ls. Science, Treatise on Mat'ls Science & Technology, 19A: 2 (1980).
26. J. B. Cohen, H. Dölle and M. R. James, "Determining Stresses from X-ray Powder Patterns", Symposium on Accuracy in Powder Diffraction, NBS Special Publication 567, p. 453 (1980).
27. H. Dölle and J. B. Cohen, "Evaluation of (Residual) Stresses in Cubic Metals", Met. Trans., 11A: 831 (1980).
28. H. Dölle and J. B. Cohen, "Residual Stresses in Ground Steels", Met. Trans., 11A: 159 (1980).

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13. ABSTRACT

The origins of residual stress and changes during fatigue are discussed. A new mechanism for fading is proposed. Practical (destructive and non-destructive) methods for measuring this stress are critically reviewed. Each technique has major problems requiring further study, but acoustic, magnetic and x-ray methods are all poised for more widespread use.

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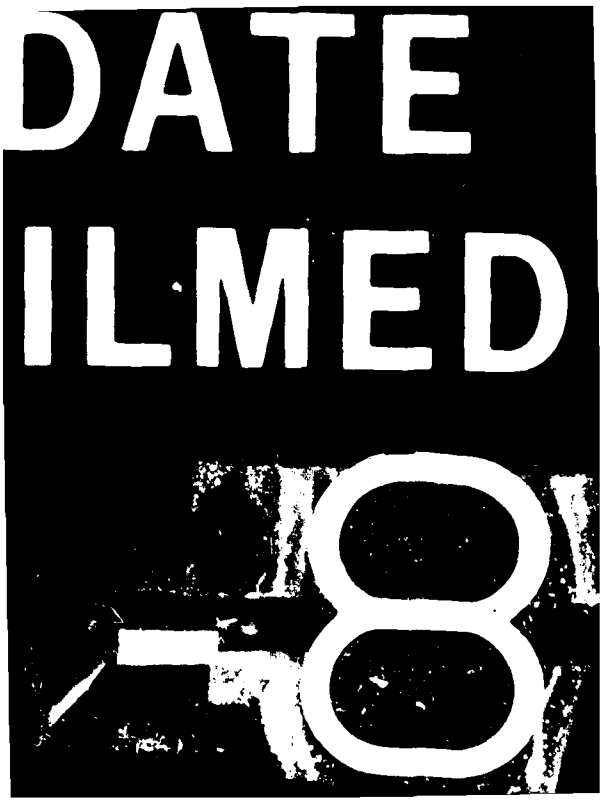
residual stress, measurements of residual stress, x-ray measurements of residual stress, fading of stresses during fatigue.						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